

Oxidizing Environment Influence on the Mechanical Properties and Microstructure of 2D-SiC/BN/SiC Composites Processed by ICVI

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(Received 11 June 1997; accepted 22 October 1997)

Abstract

2D-SiC/SiC composites have been elaborated with BN interphases of different thicknesses which were deposited on treated ¹Nicalon[®] fibers by isothermal-isobaric chemical vapor infiltration from BCl₃-NH₃-H₂ mixtures. Their mechanical behavior was investigated at 600°C in air and compared to the results obtained on similar SiC/C/SiC composites. Although these materials exhibit similar stress and strain values at rupture when loaded at room temperature, whatever the interphase, their thermomechanical resistance depends on stress type, i.e. static or dynamic. Under static fatigue the BN interphases are more efficient than the pyrocarbon (PyC) ones. Thin BN interphases tend to maintain the interfacial properties. This result could be explained by the larger microcrack distances in the tows supporting the main part of the load, according to a lower interfacial sliding resistance. In contrast the materials with a PyC interphase, which have a much higher interfacial shear resistance at room temperature, exhibit better thermomechanical behavior under dynamic fatigue at 600°C. The mechanical characteristics are related to the evolution of the fiber-matrix interfacial zone which has been studied by SEM, TEM and EELS. © 1998 Elsevier Science Limited. All rights reserved

1 Introduction

High temperature structural applications, such as gas turbine engines or space plane thermal protection systems have generated a great interest in ceramic matrix composites (CMC). These composites exhibit a combination of attractive physico-chemical properties such as high strength, toughness and

stability in severe environments. However it is now well established that the performance of chemical vapor processed composites is strongly dependent on each component of the material and on the processing parameters.^{1,2} In a recent study on the isothermal/isobaric chemical vapor infiltration (ICVI) of a boron nitride interphase, it was shown that the temperature, the gas flow and the inlet composition conditions influenced both the BN organization and the nature of the interface with the substrate.³ Therefore the knowledge and the control of each step of the composite manufacturing are of prime importance.

In order to obtain high mechanical properties, the composites generally require a compliant interphase between the fiber and the matrix.^{4,5} At first it was generally composed of a thin carbon film deposited on the fiber prior to the matrix infiltration.^{6–8} This interphase exhibits a structure that allows sliding between the fiber and the matrix, and then acts as a mechanical fuse. Nevertheless, the carbon interphase is consumed in oxidizing environments, unless it is protected with a self-protective matrix and/or an external sealing coating.^{9,10} When the whole composite is not protected, or when it is submitted to dynamic stresses that maintain the microcracks opening, the interphase is at least partially replaced by a glassy phase which strongly bonds the fiber to the matrix.^{4,11} The result is a brittle behavior of the material.

In a previous study,¹² the carbon interphase was successfully replaced by a boron nitride coating which was expected to be more oxidation resistant. The BN interphase was elaborated by ICVI from BCl₃-NH₃-H₂ mixtures at a moderate temperature of 700°C. Thus, high strength and high strain have been obtained at room temperature with

2D-SiC/SiC composites in which the fibers were chemically treated prior to any deposition stage. The surface of these treated Nicalon[®] fibers exhibited a different composition over significant thickness.¹³ However it has been shown that the interfacial sliding resistance increased with the interphase thickness at room temperature, even if a BN thickness variation in the 0.2–0.7 μm range did not induce any differences in the ultimate strength and strain. Thus crack saturation occurred for the lowest interphase thickness, and the microcracks spacing in the tensile stress direction (within the 0° fibers-tows) decreased as the BN thickness increased.¹²

The first objective of this study was to study the chemical and structural evolution of the interphase and interfaces of 2D-SiC/BN/SiC composite when exposed to mechanical stress in an oxidizing environment. It is indeed well established that the fiber/matrix interface mainly governs the thermo-mechanical behavior of such composites.^{7,14,15} Therefore, various mechanical tests have been performed such as static bending and tensile tests, as well as dynamic fatigue in tension–tension. The influence of the interphase thickness on the mechanical behavior was also examined. These mechanical characterizations were carried out in air at a temperature of 600°C which corresponds to a moderate temperature for the occurrence of hexagonal-BN oxidation. After failure of the composite, the interphase and interfaces were examined at different scales using scanning electron microscopy but also transmission electron microscopy for the microstructure and energy loss spectroscopy for the chemistry of each interfacial zone. In addition, the mechanical results obtained for the SiC/BN/SiC composites with various BN thicknesses were compared to the results obtained on similar materials with a pyrocarbon interphase.

2 Experimental procedures

2.1 Materials

The 2D-SiC/BN/SiC composites were prepared as rectangular plates (160×80×3 mm³) from 2D preforms consisting of stacks of fabrics made with Nicalon[®] fibers. These stacks were maintained and pressed together with graphite tooling to obtain a constant nominal fiber content of about 37 vol%. The Nicalon[®] fibers (NLM 202) were previously chemically treated (proprietary treatment) by the Société Européenne de Propulsion (SEP). The treated fiber presented a different composition over a significant thickness and the surface was free carbon rich.¹³ At room temperature, it has been checked that this treatment did not decrease the mechanical properties of the fibers.

The boron nitride interphase was infiltrated within the porous fiber preforms by the ICVI process. BCl₃–NH₃–H₂ gas mixtures were used at a moderate temperature of 700°C and under reduced pressure according to a procedure which has been widely described elsewhere.^{3,16} Various interphase thicknesses, in the range 0.2–0.7 μm , have been deposited. The BN-coated Nicalon[®] preforms were further densified with SiC by a classical ICVI process from methyltrichlorosilane–H₂ mixtures. Moreover, after the specimen manufacturing, a SiC coating was deposited to prevent any direct initial oxygen access to the interphase and interfaces.

Some mechanical characteristics of SiC/SiC composites that include a pyrocarbon (PyC) interphase deposited on treated Nicalon[®] fibers are provided for comparison. In all cases, the materials were overcoated by SiC after machining.

2.2 Mechanical tests

The mechanical behavior of the 2D composites was evaluated at 600°C in air by static bending and tensile tests as well as dynamic tension–tension fatigue tests. For these tests, bars were machined with their main direction parallel to one of the fiber orientation. The geometry and dimensions of the tensile and bending specimens are presented in Fig. 1. The static fatigue tests were performed on a 4-points bending machine prototype having major and minor spans of 5.08 and 2.54 cm. The tensile fatigue behavior was studied with a hydraulic machine (Instron Dynamic 8501) equipped with hydraulic grips cooled with water. These characterizations were performed within a furnace (AET) in ambient air. The tensile strain was measured with an extensometer (2620-603 Instron) with a 25 mm gauge length. For the dynamic tests in tension–tension, the frequency was adjusted at the desired value, 2 or 20 Hz in the present case, after few cycles at 0.02 Hz.

2.3 Microstructure and chemistry of the interfacial zones

The interfacial microstructures were studied by transmission electron microscopy (TEM) whereas

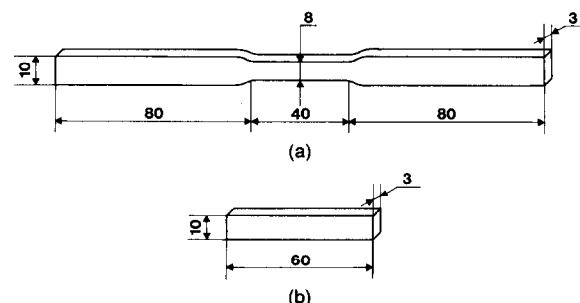


Fig. 1. Geometry and dimensions, in mm, of the (a) tensile and (b) bending specimens.

microchemical analyses were achieved by electron energy loss spectrometry (EELS). These experiments were performed on a Philips CM30ST microscope fitted with a Gatan 666 spectrometer. EELS analyses have been generally carried out with a probe size about 20 nm in diameter, reduced to 7 nm for the study of thin sublayers. The energy resolution of the spectrometer was about 1.5 eV.

The samples were taken of a tensile-tested specimen, near the failure surface. Thin foils ($\leq 500\text{\AA}$) were prepared in a conventional way by mechanical polishing followed by Ar^+ ion milling.

3 Results

3.1 Mechanical results

All the mechanical tests were performed in the non-linear domain of the composite at 600°C in air. At this temperature, a crack sealing by oxidation of the SiC matrix is unexpected.^{17,18} On the other hand, at such low temperature no decrease of the mechanical properties of the treated Nicalon[®] fiber can occur due to its thermal decomposition.^{18,19} In that way, the results are considered to be mainly representative of the evolution of the fiber/matrix interfacial zone in an oxidizing environment.

3.1.1 Static bending tests

The 2D-SiC/BN/SiC composites were made with treated Nicalon[®] fibers and various interphase thicknesses of 0.7, 0.4 and $0.2\ \mu\text{m}$. They were submitted to a static flexion fatigue testing in air at 600°C . Two specimens of each composites, including different BN thicknesses as well as a pyrocarbon interphase, were tested using a constant applied load to determine the time to cause fracture as a function of the applied stress. The values for the first specimen to fail are plotted in Fig. 2. Both samples of the composite made with a $0.2\ \mu\text{m}$ BN interphase were not fractured after more than 300 h at 150 MPa. It must be noted that the ultimate tensile strength at room temperature was about 300–320 MPa for all materials.

The static fatigue tests in air at 600°C show that the oxidation resistance of the SiC/SiC composites is greatly improved when a BN interphase is used instead of the pyrocarbon one. Moreover the failure of the composite appears later when the applied stress and the interphase thickness are low.

3.1.2 Tensile fatigue characterizations

One specimen of each material was tested in tension–tension at different stress levels for the dynamic characterizations. The results are reported as minimum time to cause failure (Table 1). The

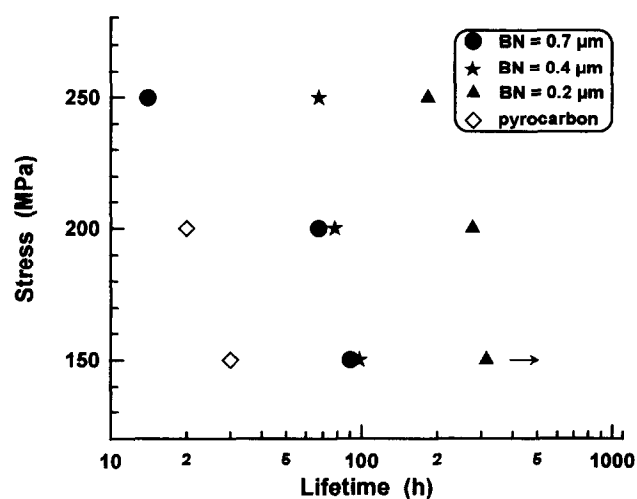


Fig. 2. Static fatigue behavior in flexion for 2D-SiC/SiC composites with various interphases deposited on treated Nicalon[®] fibers (the arrow indicates that the specimen was not fractured when the test was stopped; the pyrocarbon thickness was in the 0.1–0.2 μm range).

frequency was adjusted as a function of the applied stress. The results obtained for similar composites with a pyrocarbon interphase are also summarized in Table 1. It can be noted that various types of SiC/C/SiC composites, corresponding to different processing conditions, have been used for the characterization at 120 MPa. They broke at different time in the 15 to 24 h range. For comparison, a composite with a $0.7\ \mu\text{m}$ BN thickness did not break after about 1.6×10^6 cycles at 20 Hz (22 h) when loaded between 0 and 150 MPa, but at room temperature.

Even though only a few dynamic tests have been performed and only one specimen has been characterized at each stress level, it appears that the SiC/BN/SiC materials break before the SiC/C/SiC composites. The result at 200 MPa may be not significant because of the short duration of the test. Obviously, the composites retain their mechanical strength for a longer duration as the stress decreases. On the other hand, results for SiC/BN/SiC would suggest that lower BN thicknesses lead to better mechanical results at 600°C in an oxidizing environment. However, this behavior needs to be confirmed.

Furthermore, static fatigue tests in tension have been performed on a SiC/BN/SiC composite. In that case, the interphase thickness was $0.7\ \mu\text{m}$ and only one specimen could be loaded at different stress levels. The results are reported in Table 2. When compared with flexion tests (Fig. 2), they confirm a higher sensitivity to fatigue tests in tension than in flexion, as currently observed. If we compare with the results of Table 1 obtained in tension–tension, the results of the static tests reported in Table 2 are obviously far better.

Table 1. Minimum number of cycles and duration to cause failure of 2D-SiC/SiC composites with different interphases. They were tested in tension-tension, in air at 600°C, with various applied stresses

Stress (MPa)	Frequency (Hz)	BN thickness			PyC 0.1–0.2 µm
		0.7 µm	0.4 µm	0.2 µm	
0–120	2	—	54 000 (7 h 30 min)	68 400 (9 h 30 min)	108 000–172 800 (15–24 h)
0–150	2	9000 (1h 15 min)	—	—	19 800 (2 h 45 min)
0–200	20	22 000 (18 min)	—	—	9500 (8 min)

Table 2. Time to cause failure of 2D-SiC/BN/SiC composites submitted to static tensile stress. The interphase thickness was 0.7 µm. R indicates a rupture of the materials and NR indicates no rupture

Stress (MPa)	BN thickness 0.7 µm	
120	> 100 h	NR
145	15 h	R
185	1 h	R

3.2 Evolution of the fiber/matrix interfacial zones in an oxidizing environment

3.2.1 SEM observations

These observations have been performed on a sample made with a BN thickness of 0.5 µm which had failed after a tensile test of about 70 h at 600°C. In that case another test procedure was employed that permits to evaluate quickly the performances of a new material: the applied stress was increased by increments of 10 MPa. After each step the following increment was applied when the strain of the composite nearly reached a stable state. Finally, failure of the composite occurred at 170 MPa.

SEM micrographs of polished cross-sections show that the BN coating is uniformly deposited inside the preform [Fig. 3(a)–(c)]. The fiber/matrix interfacial zone either appears smooth or presents some crystalline or amorphous masses [Fig. 4(a) and (b)]. Microcracks are observed at the fiber/BN interface but also in the interphase itself and at the BN/matrix interface [Fig. 5(a) and (b)]. They propagate randomly in these three regions, on the contrary to the results obtained at room temperature where these cracks mainly propagate at the fiber/BN interface.¹² Many links that present a vitreous appearance are seen between BN and the fiber (Fig. 6) and between BN and the matrix. These particular glassy phases are also observed in cracks that propagate into the matrix between two neighboring fibers (Fig. 7). The fracture surface shows generally a short pull-out length [Fig. 8(a) and (b)]. Moreover, it appears that the BN coating adheres both to the fiber and the matrix [Figs 5(b) and 8(c)].

3.2.2 Microstructure and chemistry of the fiber/matrix interfacial zone

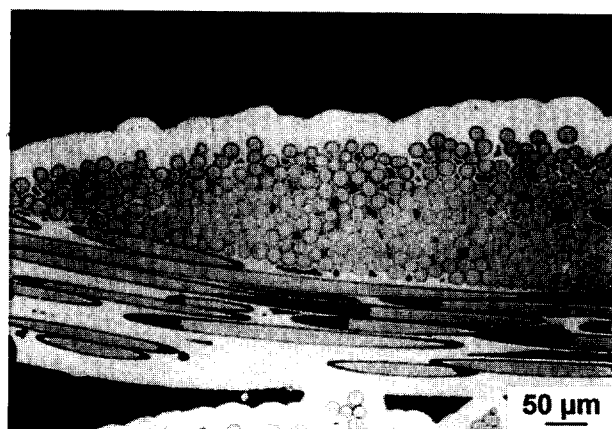
Further microstructure investigations have been made on the same sample by transmission electron

microscopy (TEM) and electron energy loss spectroscopy (EELS). The analyses were radially performed, from the fiber to the matrix through the BN interphase and its interfaces both with the treated Nicalon[®] fiber and the SiC matrix, according to a procedure which has been described elsewhere.¹⁵ It is interesting to recall that the interfaces were initially carbon-rich, before any exposure to the oxidative environment.¹⁵ Because our interest was principally focused on the chemistry of the interfacial zone, no attempt was made to characterize the matrix and the fiber.

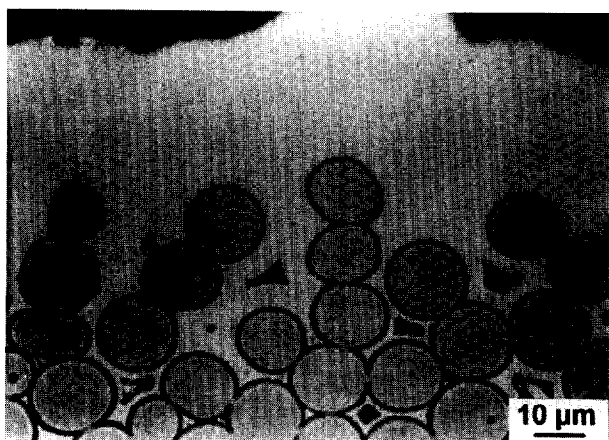
3.2.2.1 The BN interphase. The thickness of the BN interphase is homogeneous (about 0.5 µm) and no voids are observed (Fig. 9). The EELS spectra of this region show boron and nitrogen edges (Fig. 10) typical of these elements involved in hexagonal boron nitride.²⁰ In addition, they evidence significant amounts of carbon and oxygen in the BN phase. Silicon was not detected, however it could be below the detection limit of EELS (<1 at%). The concentration of B, C, N and O was determined by a classical quantitative analysis technique.²¹ The average atomic composition was 40% N, 48% B, 7% C and 5% O. In comparison with the average composition measured on the same composite tested at room temperature (41.5% N, 39% B, 10% C and 9.5% O),¹⁵ we note a decrease of the O, C and N contents.

3.2.2.2 The fiber/BN interface. The TEM micrographs show many large decohesions at this interface as clearly shown in Fig. 9. Otherwise, when the fiber and the interphase are still bonded, a bright zone of about 10 nm thick borders the fiber. This sublayer is mainly composed of silicon and oxygen with traces of boron and carbon. The near-edge fine structure of the Si-L_{2,3} edge (double peak at 109 and 116 eV) (Fig. 11) and the dissymmetrical plasmon peak at 22.5 eV in the range of low energy loss (Fig. 11, inset) are consistent with the presence of silica at this interface. The quantitative analyses give an average Si/O atomic ratio of 0.53. It can be recalled here that the treated Nicalon[®] NLM 202 fiber presented a carbon-rich surface which was maintained at the fiber/BN interface, at the end of the composite manufacturing.¹⁵

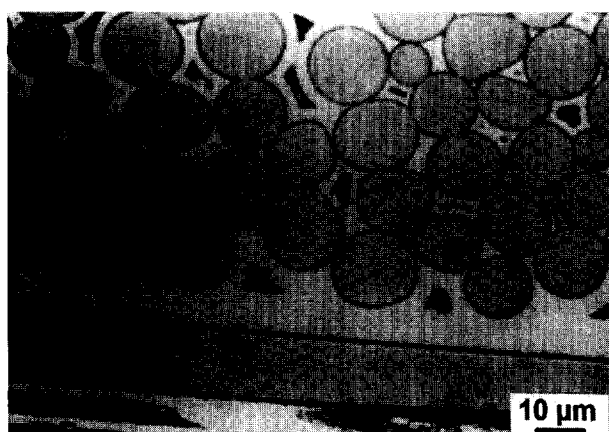
3.2.2.3 The BN/SiC-matrix interface. Figure 9 shows the interface between the SiC matrix and the BN interphase. It is coarse and porous with a thickness of about 35 nm. Although the feature of this zone appeared similar to the interface observed on the same composite tested at room temperature, the composition is significantly modified. EELS analyses have revealed high concentration of oxygen (50 at%), silicon (34 at%) and carbon (16 at%), instead of carbon in the original composite.¹⁵ Additionally, the damped two pre-peaks in



(a)



(b)



(c)

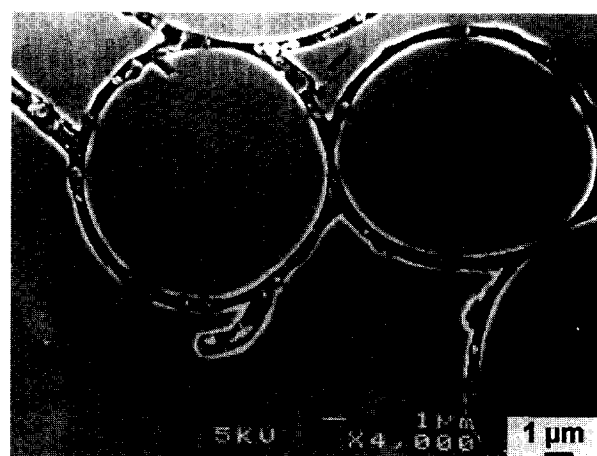
Fig. 3. (a) Cross-section of a 2D-SiC/BN/SiC composite, (b) in the external part and (c) in the middle of a fibers tow.

the near-edge fine structure of the Si-L_{2,3} edge may suggest that Si atoms are involved in both SiC and SiO₂¹⁵ (Fig. 12). An additional HREM study could permit to afford more information about the relative organization of SiC, SiO₂ and C. Furthermore, close to this interface, inside the boron nitride, some silica has also been detected.

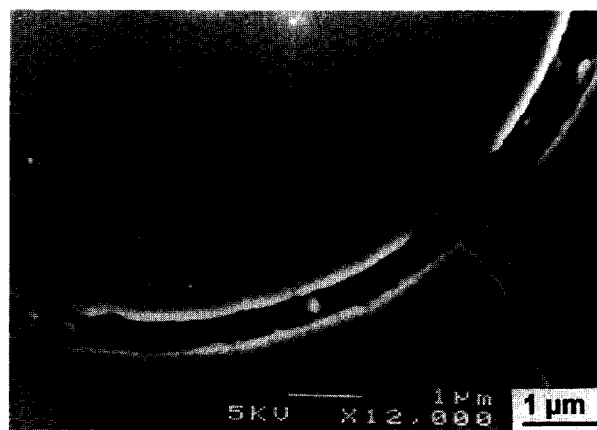
3.2.2.4 Analysis of the cracks. Along a decohesion at the fiber/BN interface, EELS analyses revealed the presence of Si, O and B atoms. Carbon was also revealed, but it remains under the quantitative limit, while no nitrogen was detected. The average atomic ratio O/Si was equal to 1.3 and the boron concentration was below 5 at%.

On the other hand, the chemical characterization of a crack that propagates radially into the interphase allows to determine an average atomic composition of 34% B, 28% N, 20% O, 10% C and 8% Si. The near edge structure of the Si-L_{2,3} peak exhibits a double peak on the edge onset. It is similar to the one previously seen at the fiber/BN interface and corresponds to the presence, in this crack, of Si-O bonds in a SiO₂ structure.²²

Although, in both cases, the presence of boron oxide could be considered from the surstoichiometry



(a)



(b)

Fig. 4. Interfacial zone showing crystalline or amorphous masses after a tensile test of 70 h, at 600°C in air.

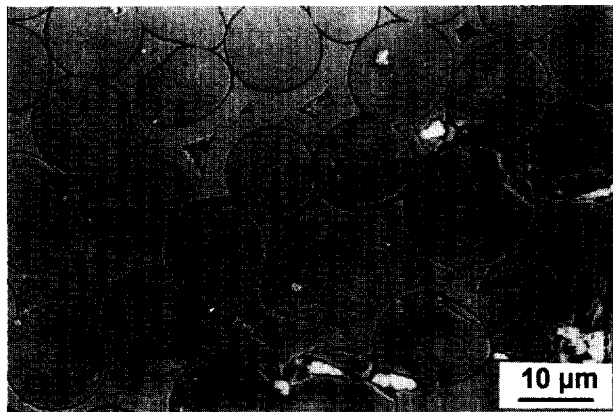
in boron as compared with nitrogen, that cannot be evidenced by our analyses.

4 Discussion

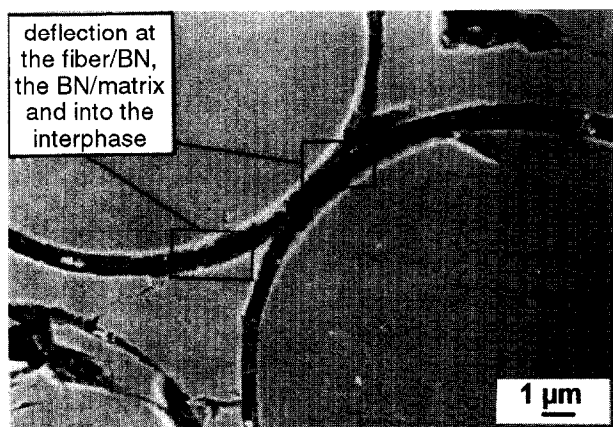
The application of a compliant interphase such as carbon or boron nitride on treated Nicalon[®] fibers has been found to improve the mechanical behavior of the SiC/SiC composites at room temperature. The drawbacks of carbon are obvious in oxidizing atmosphere, and BN should exhibit a better oxidation resistance, especially at moderate temperature such as 600°C. However the ICVI–BN deposits always present a turbostratic structure. Even if the temperature of 600°C is low for the onset of hexagonal-BN oxidation, poorer lattice organization might be affected at this temperature. The availability of boron oxide could then allow the formation of a glassy phase with various amounts of boron. It is indeed well-known that boron significantly decreases the viscosity of glasses. The melting temperature of B₂O₃ is about 480°C²³ and the eutectic temperature of the pseudo-binary SiO₂–B₂O₃ is, for example, about 450°C.²⁴ When considering the flowing ability and

the volatility of boron oxide, it is not surprising to find such a phase in different parts of the composite, especially in the tows, inside cracks perpendicular or parallel to the fibers (Figs 6 and 7). Sometimes it can even link the cracks walls. The EELS analysis along a decohesion at the fiber/BN interface confirms that the Si, O and B elements are present (Section 3.2.2, Analysis of the cracks). The analysis of a crack inside the BN interphase reveals also an increase in oxygen content and the silicon presence with Si–O bonds formation.

From EELS study, it appears that the bulk boron nitride interphase can be modified at 600°C in air. However the decrease of the O, C and N contents should be examined by further analyses, particularly at different distances from the far end of the cracks, i.e. different conditions of ambient air access. This modification could be attributed to the relatively poor organization of the structure and the initial incorporation of C and O impurities. Because its composition was not changed by the ICVI processing of the silicon carbide matrix from



(a)



(b)

Fig. 5. Cracks deflection in a 2D-SiC/BN/SiC composite tested in tension at 600°C in air.

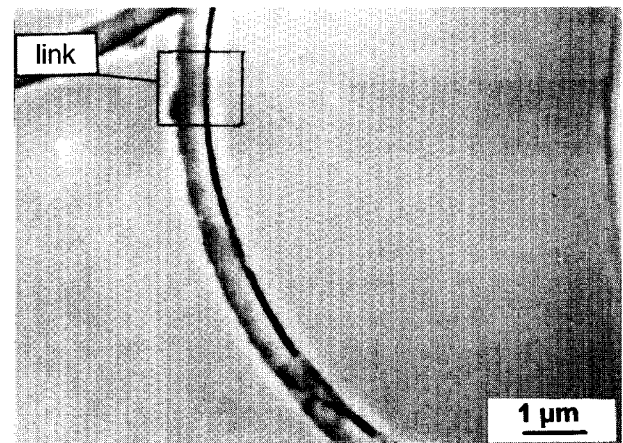


Fig. 6. Vitreous phase linking the treated Nicalon[®] fiber and the BN interphase.

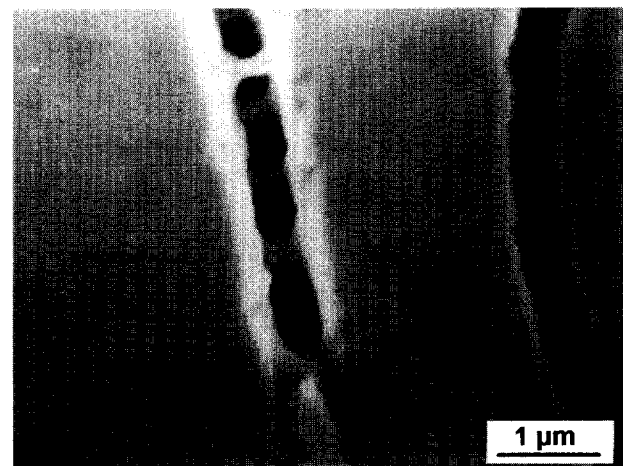


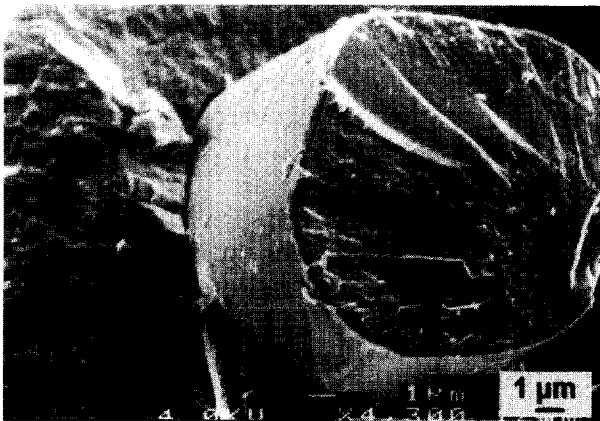
Fig. 7. Glassy phase in a crack propagating between two neighboring fibers.

a hydrogen–methyltrichlorosilane mixture at about 1000°C,¹⁵ these variations might be due to some reorganization resulting from an initial interaction between oxygen from ambient air and C impurity. Beyond the far end of the cracks, this boron nitride layer could be one way for oxygen access at a low rate inside the composite, especially to the interfaces with the fibers and the matrix. As shown in the following, the silica formation in place of free carbon should significantly decrease the oxygen

access parallel to the fibers, in these very thin sub-layers (oxygen diffusion through silica is very slow; it increases with the boron content in glasses but the amount of boron is low in places that do not present microcracking).



(a)



(b)



(c)

Fig. 8. Fracture surface of a 2D-SiC/BN/SiC tested in tension at 600°C in air. Rupture occurred at 170 Mpa after 70 h.

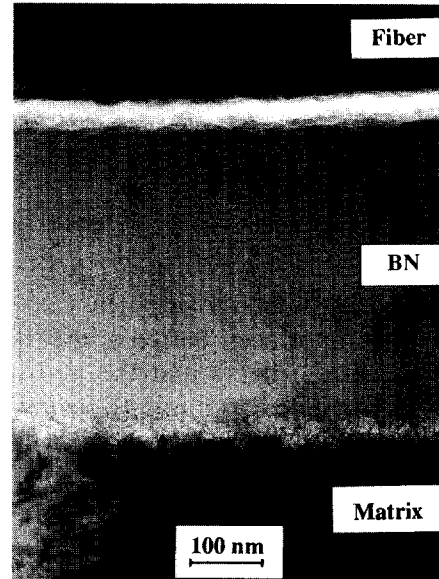


Fig. 9. Micrograph of the BN interphase and its interface with the treated Nicalon[®] fiber and with the SiC matrix.

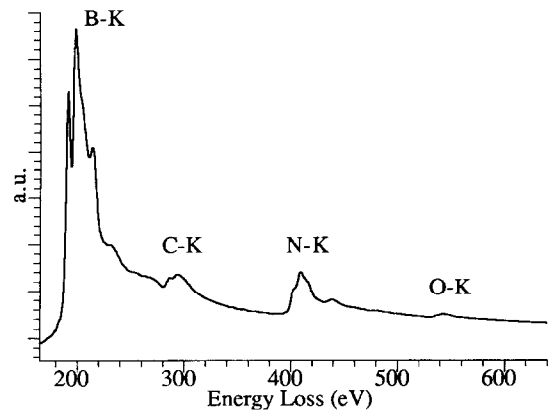


Fig. 10. EELS spectrum of the BN interphase.

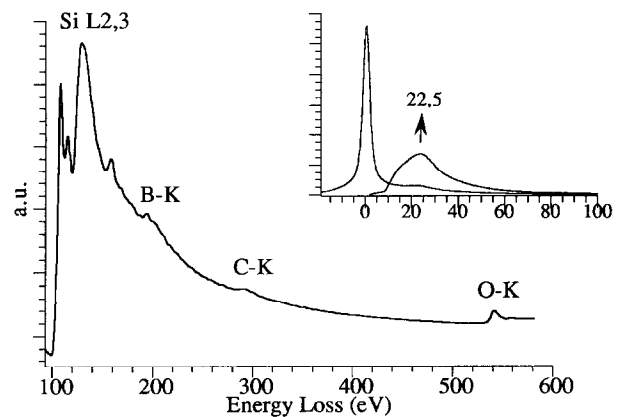


Fig. 11. EELS and plasmon (inset) spectra of the treated Nicalon[®] fiber/BN interface.

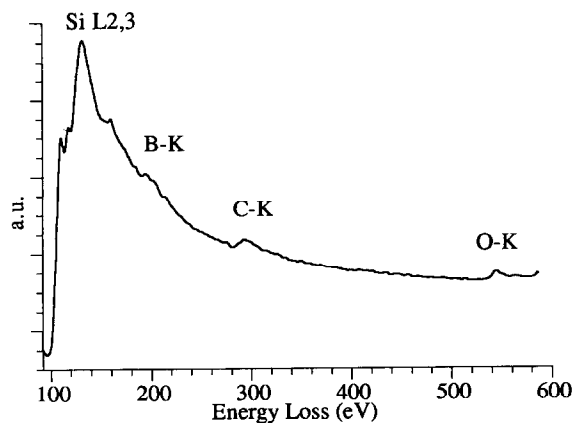


Fig. 12. EELS spectrum of the BN/SiC-matrix interface.

It is indeed particularly important to examine accurately the interfaces between the BN interphase and both the fibers and the matrix. Crack deflection from mode I to mode II occurs either at the fiber/BN or BN/matrix interfaces where carbon is the main initial constituent, as shown previously.¹⁵ The deflection mechanism at the later interface was not (or rarely) observed at room temperature owing probably to a weaker fiber/BN link and different stresses field. When the BN interphase is still bonded with the fibers and the matrix, the initial carbon sub-layer has been consumed with the formation of silica at the BN/fiber interface. At the BN/SiC interface, the initial free carbon has been oxidized or transformed in silica and silicon carbide. These silica or silica plus silicon carbide formation clearly leads to strong interfacial bondings. When compared to room temperature properties, a worse mechanical behavior can therefore result.

Obviously, the evolution of the interfacial zones at a given temperature depends on time but also on the ability for oxygen to reach the oxidizable layers. At this temperature, a sealing of the SiC/matrix cracks is considered to be negligible, especially far from the BN interphase and when a dynamic loading is applied. Thus, the failures of the composites appear all the later since the applied stress is low as the total crack opening is directly correlated to strain. On the other hand, the total amount of crack opening depends on both cracks width and cracks density. A higher density of microcracking might increase the oxidation rate of the interfacial sublayers if we suppose that the gaseous diffusion within the radial cracks is not rate-limiting (a reasonable hypothesis at such moderate temperature of 600°C). In that case, more interfacial zones are damaged when the cracks density increases. Because the crack density decreases when the BN interphase thickness decreases (at least in the range 0.2–0.7 μm) the better behavior under static loading of the composites that include

the thinnest BN interphase could be explained. For the same reasons, also when considering the very thin carbon sublayers (when compared to the pyrocarbon interphases) and some possible protection by oxide formers, SiC/BN/SiC composites are more oxidation resistant than SiC/C/SiC materials under static loading (Fig. 2).

In contrast, the composites with a pyrocarbon interphase exhibit higher resistance to dynamic fatigue in tension–tension, at 600°C in air. In that case, the mechanical behavior depends on the sliding and wear resistance, besides the stability of each constituent in the oxidizing environment and the ability for oxygen to diffuse into the composite. Even if it seems that the materials with a BN interphase are less altered by oxidation, it has been shown¹² that the SiC/C/SiC composites present a better dynamic fatigue resistance at room temperature. This has been related to the much higher interfacial shear resistance, twice to five times greater than those of materials with a BN interphase. With regard to the influence of the BN thickness, thinner interphases lead to lower sliding resistance and less developed microcracking inside the 0° tows at room temperature. Then, it could be supposed that the fiber/matrix bonding changes slower during mechanical tests at 600°C than those of thicker interphases. The result is a better retention of the initial mechanical properties. These hypotheses are based principally on characteristics evaluated at room temperature, especially the sliding resistance and the crack spacing. A more thorough study should consider the coupled influence of the temperature with the interphase thickness, the influence of the stress field in the fiber/matrix region, and the environmental effects on the interfacial properties.

5 Conclusion

The large potentialities of ceramic composites require the preservation of their thermomechanical properties for long duration, in oxidative environments, under high temperature fatigue conditions. This implies that the fibers, the matrix and the interfacial properties such as load transfer, wear resistance and crack deflection could be maintained.

Boron nitride interphases appeared to be more efficient than pyrocarbon ones under static fatigue at 600°C in air. Although boron nitride, deposited by ICVI from BCl₃–NH₃–H₂ mixtures, seems to be partially oxidized even at this moderate temperature, its resistance towards corrosion is still higher than carbon. As a result of this oxidation, a Si–B–O glassy phase has been found in the cracks

propagating through the material. The decrease of the mechanical properties at high temperature is mainly attributed to the evolution of the carbon-rich interfaces leading to the main formation of silica and consequently to a stronger fiber/matrix bonding. This evolution can be slowed down by controlling the access of oxygen till the oxidizable layers. Thinner BN interphase decreases the interfacial shear resistance and mainly the crack density. Thus, interfacial properties of materials with small BN thicknesses, even if they are initially lower, are less changed and a better thermomechanical resistance to static and dynamic constraints is obtained. The classical SiC/C/SiC materials which have a much higher interfacial shear resistance still exhibit a better behavior under dynamic fatigue at 600°C in air, while being more oxidizable. It appears then of prime importance to increase greatly the interfacial sliding resistance of SiC/BN/SiC composites to improve both the static and the dynamic behaviors of such materials. On the other hand, it would be interesting to prevent the formation of carbon-rich interfaces resulting from the fiber treatment and from the SiC/matrix infiltration.

Finally, further studies should consider the influence of the temperature on the interfacial properties and the damaging modes of the materials. If the oxygen access parallel to the fibers through the BN interphase itself is important, the BN purity, its organization and oxidation resistance should be improved, and multi-sequenced interphases that include thin BN sub-layers should be studied.

Acknowledgements

This research has been supported by the Société Européenne de Propulsion and by the Région Centre through a grant given to M. Leparoux. The authors acknowledge J. Gouineau and G. Lama-zouade from SEP for the mechanical characterizations and for the SEM observations, respectively.

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